AD-A098 802 CARNEGIE-MELLON UNIV PITTSBURGH PA DEPT OF METALLURG--ETC F/G 11/6
THE ROLE OF HYDROGEN IN THE STRESS CORROSION CRACKING OF HIGH S--ETC(U)
MAR 81 A W THOMPSON, I M BERNSTEIN F49620-77-C-0003
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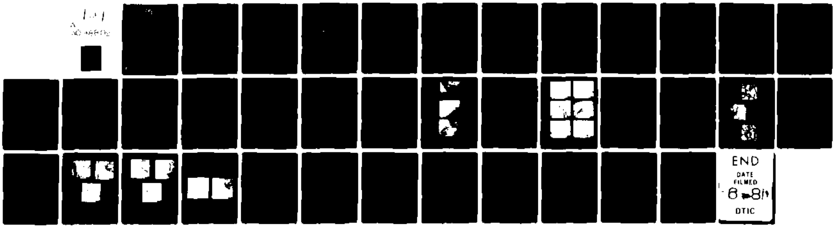

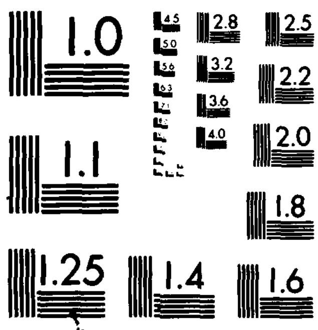

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The Role of Hydrogen in the Stress Corrosion Cracking of High Strength Aluminum Alloys.

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REPORT DOCUMENTATION PAGE		READ INSTRUCTIONS BEFORE COMPLETING FORM
1. REPORT NUMBER AFOSR-TR- 81-0403	2. GOVT ACCESSION NO. AD-A098802	3. RECIPIENT'S CATALOG NUMBER
4. TITLE (and Subtitle) "The Role of Hydrogen in the Stress Corrosion Cracking of High Strength Aluminum Alloys:		5. TYPE OF REPORT & PERIOD COVERED Final: 1 December 1976 - 31 December 1980
		6. PERFORMING ORG. REPORT NUMBER
7. AUTHOR(s) A. W. Thompson, I. M. Bernstein		8. CONTRACT OR GRANT NUMBER(s) F49620-77-C-0003
9. PERFORMING ORGANIZATION NAME AND ADDRESS Department of Metallurgical Engineering and Materials Science Carnegie-Mellon University, Pgh. PA 15213		10. PROGRAM ELEMENT, PROJECT, TASK AREA & WORK UNIT NUMBERS 61102F 2306/A1
11. CONTROLLING OFFICE NAME AND ADDRESS AFOSR/NE, Bldg. 410 Electronic & Solid State Sciences Bolling AFB, Washington, D.C. 20332		12. REPORT DATE 1st March, 1981
		13. NUMBER OF PAGES 33
14. MONITORING AGENCY NAME & ADDRESS (if different from Controlling Office)		15. SECURITY CLASS. (of this report) UNCLASSIFIED
		15a. DECLASSIFICATION DOWNGRADING SCHEDULE

16. DISTRIBUTION STATEMENT (of this Report)

Approved for public release; distribution unlimited.

17. DISTRIBUTION STATEMENT (of the abstract entered in Block 20, if different from Report)

18. SUPPLEMENTARY NOTES

19. KEY WORDS (Continue on reverse side if necessary and identify by block number)

Hydrogen embrittlement, stress corrosion cracking, aluminum alloys, microstructure

20. ABSTRACT (Continue on reverse side if necessary and identify by block number)

The objective of this program has been to understand the stress corrosion cracking (SCC) behavior of high-strength aluminum alloys, and in particular to understand the role played by hydrogen embrittlement in such cracking. The approach taken was to study microstructural effects on both hydrogen embrittlement and SCC, and to establish, insofar as possible, microstructural and fractographic correlations with cracking behavior, and detailed understanding of the mechanical behavior of hydrogen-charged material.

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When this program began, there existed only a single unconfirmed report of hydrogen embrittlement in a high-strength aluminum alloy. That report was confirmed (on 7075-T651), and extensive additional evidence acquired (on 7075, 2124 and 7050, in a variety of heat treatments), permitting us to assemble a complete and detailed description of hydrogen embrittlement and its dependence on microstructure in Al alloys, in some 14 publications.

A parallel effort on SCC behavior has also been conducted, using the technique of Mode I- Mode III testing to provide information on the relative contributions of hydrogen embrittlement and anodic dissolution to SCC. The Mode I data for 7075 showed a close parallel to the hydrogen charging results, in microstructural dependence and in fractography. The Mode III data showed that the anodic dissolution component of SCC is small compared to hydrogen embrittlement, and moreover, has a much weaker dependence on microstructure.

9 FINAL TECHNICAL REPORT. 1 Dec

The Role of Hydrogen in the Stress Corrosion Cracking of High Strength Aluminum Alloys

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CMU Report Number: AFOSR-A1-4

Grant Number: [redacted] F49620-77-C-0003

Period: 1 Dec. 76-31 Dec. 80

Principal Investigators: A. W. Thompson, I. M. Bernstein

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1.0 ABSTRACT OF RESULTS

The objective of this program has been to understand the stress corrosion cracking (SCC) behavior of high-strength aluminum alloys, and in particular to understand the role played by hydrogen embrittlement in such cracking. The approach taken was to study microstructural effects on both hydrogen embrittlement and SCC, and to establish, insofar as possible, microstructural and fractographic correlations with cracking behavior, and detailed understanding of the mechanical behavior of hydrogen-charged material.

When this program began, there existed only a single, unconfirmed report of hydrogen embrittlement in a high-strength aluminum alloy. That report was confirmed (on 7075-T651), and extensive additional evidence acquired (on 7075, 2124 and 7050, in a variety of heat treatments), permitting us to assemble a complete and detailed description of hydrogen embrittlement and its dependence on microstructure in $A\ell$ alloys, in some 14 publications.

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2.0 TECHNICAL RESULTS

The intent of the program was to identify the role played by hydrogen in stress corrosion cracking by establishing microstructural and fractographic correlations with cracking, and by mechanical tests under cathodic charging conditions. Specific emphasis was placed on the role of such metallurgical variables as precipitate type, size and distribution, in order to develop them as control variables for the production of high strength aluminum alloys more resistant to hydrogen embrittlement (HE) and/or stress corrosion cracking (SCC).

2.1 Hydrogen Embrittlement of Commercial 7075 - T651

The first topic of this program was to study in detail the response of commercial 7075 aluminum alloy material to cathodic charging, with emphasis on a critical examination of the Gest and Troiano results (1). The cathodic charging procedure adopted was as follows: specimens were charged for ten hours under potentiostatic control in an HCℓ solution of pH 1 with an applied potential of -1500 mV versus standard Calomel electrode.

Tensile tests were performed on these charged specimens and also on uncharged specimens, over the temperature range from -98°C to 20°C. The ductility of the charged specimens was significantly lower for all testing temperatures (Figure 1). The degree of intergranular secondary cracking, was dependent on both test temperature and charging conditions. It was therefore shown by these tests that there is clearly a hydrogen effect, manifested both as a loss in ductility and as an increase in secondary cracking (2). This confirms and extends the results of Gest and Troiano (1).

2.2 The Role of Microstructure in HE of Commercial 7075 Aℓ

The second year's work concentrated on the role of microstructure in the hydrogen embrittlement process. The microstructural variable that is thought to be of primary importance in the environmental fracture behavior of aluminum alloys is the character (nature and distribution) of the precipitate population (3,4). The influence of precipitates both in the grain interior and at grain boundaries can be explained by two behavior features: (i) the character of those grain interior precipitates which control slip mode (3,5); (ii) the relative coverage of grain boundaries by the intergranular precipitates. The experimental techniques used for examining the role of the precipitates have been described in detail (6).

In summary, the results were as follows (6). The response of 7075-aluminum having three different microstructures (underaged, peak-aged and overaged), Figure 2, to cathodically-charged hydrogen was examined in the temperature range of -200°C to 20°C. The principal results were:

- For all microstructures, a temperature dependent loss of ductility due to hydrogen was found, Figure 3.
- The effect was absent at the lowest test temperature, ruling out a purely mechanical effect due to charging damage present prior to mechanical testing.
- The temperature dependence of the embrittling effect showed a definite maximum, which was dependent on the microstructure; with increasing aging temperature, the maximum shifted towards lower test temperatures.
- The ductility loss was a function of microstructure, being highest for the underaged and lowest for the overaged material, which suggests a correlation between hydrogen embrittlement and slip planarity. The pattern found qualitatively here mirrors the pattern found for stress corrosion cracking susceptibility.
- It was found that the fracture mode was not affected by microstructure and charging conditions (Figure 4). For all temperatures, a transgranular dimpled type fracture was observed, which is believed to be due to the test geometry relative to the pancake-shaped grains. The

dimple size was approximately the same for charged and uncharged material, which suggests enhanced nucleation and growth of microvoids due to hydrogen. This conclusion is reached since with the smaller RA in hydrogen charged material, the dimples would be expected to be larger, purely on geometrical grounds.

2.3 Strain-Rate Dependence of HE in Commercial 7075 Aℓ

Having established that the embrittlement of 7075 aluminum by cathodically charged hydrogen is strongly dependent on microstructure, the strain rate dependence was then investigated. The results (Figure 5) showed that an increase in tempering temperature decreases the sensitivity to hydrogen embrittlement at the strain rate of maximum effect. The fact that UT is the most susceptible to hydrogen embrittlement, and T73 least susceptible, can be rationalized in terms of the microstructure and the resultant slip, which becomes progressively finer and more homogeneous with increased degree of tempering (7).

2.4 HE of High Purity 7075

To further investigate the role of precipitates and slip distribution, additional tests were begun in the 2nd year of the investigation, using a higher purity laboratory version of the 7075 alloy; this can be heat treated to have an equiaxed grain structure. Whereas the commercial alloy did not show any dramatic change in fracture mode due to cathodically charged hydrogen, a transition from transgranular fracture to a brittle, intergranular fracture was found for the high purity alloy in the underaged condition. In the underaged condition, this alloy forms intense slip bands, whereas in overaged material the slip distribution is homogeneous (8). Since there is evidence that the efficiency of dislocation transport of hydrogen is strongly correlated to slip planarity, it was hoped that appropriate experiments would yield information about the presence and efficiency of any transport process.

A straining electrode test (SET) procedure was adopted in which the specimen is deformed plastically during simultaneous charging. With this procedure, the specimens were preloaded to approximately 70% of the yield stress, and then cathodically charged in a pH 1 HCℓ solution under potentiostatic conditions ranging from -1500mV to -2000mV while being plastically strained at a very low strain rate, e.g. $4 \times 10^{-5}\ s^{-1}$. Charging was carried out from 60 to 120 minutes with a concomitant plastic strain of a few percent. The specimens were then tested at normal strain rates to fracture. If dislocation transport were the controlling process in HE, a drastic effect would be expected, resulting in a large ductility loss and a widely increased zone of intergranular fracture (especially for the underaged condition).

For commercial-purity 7075-UT, both of these charging procedures resulted in a similar RA loss due to hydrogen of about 45%. In the HP 7075-UT, the loss due to precharging was also about 45%, but for the SET it was over 90%, a very severe embrittlement, Figure 6.

Fractographic studies paralleled the above observations, Figure 7. The precharged specimens showed a small amount of intergranular fracture near the surface. This intergranular region was considerably extended by the SET procedure and, in fact, the prestraining time could be adjusted to result in a fracture which was almost completely intergranular. In the commercial 7075-UT alloy, by contrast, the precharged and SET fracture surfaces were identical and similar to the precharged HP 7075-UT results (9).

We believe these results to be very important and unique. Not only have we demonstrated that a 7075-type alloy can be completely intergranularly embrittled, but we have further shown that the extent and severity of the process is controlled by the dislocation transport of hydrogen. Similar processes probably occur in 'pancake' structures typical of the commercial composition, but are less serious (at least in the longitudinal direction) due to the grain shape effect.

2.5 Stress Corrosion Cracking of 7075 - Mode I/Mode III Testing

Having obtained an understanding of the hydrogen embrittlement behavior of 7075, the emphasis in the third and fourth years shifted to stress corrosion cracking (SCC) tests on the same microstructures. The procedure used was to vary loading mode, a concept which is believed to furnish a means of identifying the proportion of SCC susceptibility which arises from a hydrogen contribution (10). Because of the difference in stress state at the crack tip for the two modes, hydrogen's role in the fracture process can be more easily identified.

Two types of specimens were used for these tests, short transverse and longitudinal in orientation (11,12). Specimens in the short transverse (thickness) direction were machined from the previously used 7075 plates. After machining, the short transverse specimens were mechanically polished to a mirror finish to ensure a consistent surface condition for the SCC testing. These specimens had the same gage length and gage diameter, and were taken from the same plate as the specimens used in the cathodically-charged hydrogen tests (6) to ensure a common basis for comparison. Specimens in the longitudinal direction were also machined from the same plate, to provide a comparison to the earlier work (6,7) on this orientation, and circumferentially notched to permit a more accurate characterization of the initial stress state. All specimens were solution heat treated at 465°C for 20 minutes and then quenched into ice water. The short transverse specimens were solutionized after mechanical polishing to anneal out any mechanical damage due to the polishing, but the longitudinal notched specimens had to be solutionized before notching to prevent warpage during heat treatment.

The specimens were then pre-aged for at least two days at room temperature. One of three final aging treatments was used: T6 (24 hours at 120°C), T73

(24 hours at 163°C), or an underaged temper, UT (24 hours at 100°C).

SCC testing was done at a constant load (Mode I) or constant torque (Mode III) in a 50°C, 3.5% NaCℓ salt solution. Time-to-failure results for each microstructure for the short transverse specimens are shown in Figure 8. These results show the well-known pattern in Mode I, that the most susceptible SCC condition is the underaged condition. Slightly less susceptible to SCC is the T6 temper and the most resistant to SCC is the overaged temper. Of greater interest is the comparison of Mode I and Mode III results in Figure 8. Mode III clearly caused much longer failure times at a given stress and apparently also a distinctly higher "threshold stress", although much longer test times would be needed to verify the latter point.

Figure 9 shows SCC results for notched longitudinal specimens. The ordinate shows the fraction of the 50°C failure fracture toughness in oil to which the specimen was originally loaded. The toughness ratios for the notched longitudinal specimens show the same trends at the stress ratios for the smooth short transverse specimens. That is, the SCC resistance increases as aging goes from UT to T6 to T73.

The SET procedure was conducted with notched specimens in Mode I loading in pH 1 HCℓ solutions at room temperature with a cathodic overpotential of 150 mV. Crosshead rate was about 10^{-4}cm/s. The SET results showed toughness values similar to those for specimens tested in air for both the T6 and T73 tempers, as shown in Table I. The fracture toughness value for the UT temper, however, dropped about 32% in the SET. This suggests more embrittlement in the UT temper than in either the T6 or T73 tempers.

This result seems to coincide with the constant load results, which showed that the UT temper is more susceptible to hydrogen than either the T6 or T73 tempers.

Table I

Fracture Toughness Values for Notched Longitudinal Specimens

	Air	SET
UT	24.3	16.6 MPa $\sqrt{m}$
T6	25.2	25.3
T73	23.1	23.1

Fractography of the short transverse tensile specimens is shown in Figure 10. Each temper showed flat, cleavage-like regions around the circumference, with a ductile overload region in the center. Figure 11 shows similar fractographic appearances for each temper tested in tension in the longitudinal orientation. The SET specimens had similar fracture surfaces but with a wider flat region (~300 μm for SET in tension vs. ~200 μm for constant load in tension) plus more secondary cracking, as shown in Figure 12.

It has been pointed out (12,13) that a nominal Mode III loading of a real material is likely to produce a small hydrostatic component of stress near the crack tip. This fact, however, should not obscure the main issue: that the Mode III hydrostatic component is greatly reduced (whether or not it is zero) relative to Mode I. Consequently, there appears to be no error in evaluating a difference in SCC sensitivity between Modes I and III as reflecting a <u>relative</u> contribution of hydrogen embrittlement to SCC. It should be noted that the extent of this relative contribution is minimized, because the small hydrostatic stress in the real Mode III test presumably assists a correspondingly small hydrogen contribution to SCC. If a pure Mode III test could be conducted, it follows that the difference between Modes I and III would increase. Figure 8 thus indicates a minimum or conservative extent of the contribution of hydrogen embrittlement to SCC of 7075 aluminum in the UT, T6 and T73 tempers.

There are two important consequences of this phase of the work. First, the results obtained by Green, et al., (10) have been extended to a commercial material of wrought grain structure, in several heat treatments. Secondly, the apparent relative contribution of hydrogen in the SCC of the different tempers closely follows the effect of internal hydrogen on RA loss (6,7), suggesting that non-hydrogen contributions such as anodic dissolution are much weaker functions of microstructure than is hydrogen embrittlement.

2.6 Hydrogen Embrittlement of 2124 Aℓ

Work was begun in the third year of the program and continued in the fourth on an alloy of the 2000-series (based on Aℓ-4% Cu). The 2000-series alloys share one important behavioral characteristic with their high-strength 7000-series counterparts: both are susceptible to SCC (3-5). Consequently, as an extension of the work on the 7000-series alloys, the susceptibility to hydrogen of a 2000-series alloy, 2124 (Cu-4.8%, Mg-1.46%, Fe0.74%, Mn-0.5%, Si-0.099%, Ti-0.19%, Ni-0.01%) was examined. It was expected that the microstructural understanding gained from the 7000-series work could be extended to these 2000-series alloys.

Longitudinal specimens were cut from 2124 plate and heat treated to give the following standard tempers (14):

Temper:	Treatment:	
T4	50 hrs/RT	Naturally Aged
T3	2% strain + 50 hrs/RT	
T6	9 hrs/190°C	Artifically Aged
T8	2% strain + 9 hrs/190°C	

In addition, the following non-standard aging treatments were also employed:

Treatment:	Designated here as:
2% strain + 50 hrs/RT + 3 hrs/190°C	T3+3
4 hrs/190°C	UT
24 hrs/190°C	OT

Precharging with hydrogen, in a manner similar to that employed for 7075, did not cause a measurable loss in ductility. For the SET procedure, however, (similar to that used on HP 7075, above), the results are tabulated below.

Table II

Tensile Results for 2124 Aluminum

Temper	$\sigma_{0.2}$(ksi)	Mean RA % a) in air	b) SET	% RA loss
T4	39	33	27	24
T3	47	29	22	23
T6	51	25	25	0
T8	65	30	25	17
T3+3	50	30	30	0
UT	42	24	18	21
OT	46	31	31	0

Although this alloy is less sensitive to hydrogen than 7075, there is a measurable ductility drop once sufficient hydrogen absorption has occurred, as can be accomplished with the SET procedure. Little evidence was found during fractographic studies of an influence of hydrogen in the fracture mode. In all cases, the orientation of the stress axis with respect to the pancake grain shape (similar to that of 7075) was evidently the overriding influence, since a transgranular dimple-type fracture was observed in all cases.

In general, the results can be understood in terms of the framework developed from experience with the 7075 alloys (6-9). The specimens whose microstructure was characterized by the fine homogeneous distribution of coherent precipitates i.e., T3, T4 and UT exhibited RA losses on SET hydrogen charging while those in which the microstructure was closer to equilibrium, in this case the semi-coherent S' ($A\ell_2CuMg$), no RA loss was detected. One exception to this was found in the T8 temper, which exhibited hydrogen embrittlement while belonging to the latter group: this result was inexplicable despite a TEM investigation of the microstructure (14).

2.7 Hydrogen Embrittlement of 7050 Aℓ

It is well known that the addition of copper to high strength Aℓ-Zn-Mg alloys simultaneously improves their yield strength and stress corrosion cracking resistance (5). An overaging heat treatment may be effectively used to greatly enhance stress corrosion resistance without unacceptable strength penalties in copper-containing alloys.

Hence, in the fourth year of the program, a study was begun on the influence of copper additions on the hydrogen embrittlement of 7000 type high strength Aℓ-Zn-Mg alloys. The influence of copper on microstructure, aging kinetics and dislocation structure or deformation mode when correlated with resistance to embrittlement by cathodically charged hydrogen will provide information on the role of copper.

To date, testing on one copper containing alloy, 7050 (Aℓ-2.11 Mg-5.97 Mg-2.11 Cu-0.06 Si-0.07 Fe-0.02 Ti-0.12 Zr) has been completed and, as shown in the table below, this alloy is not embrittled by cathodically charged hydrogen in the overaged, T7351, condition and only exhibits a small effect in the peak aged, T651, condition.

X7050 - Longitudinal, % RA

($\dot{\varepsilon}_i = 10^{-5}\ \text{sec}^{-1}$)

	Uncharged	Charged
T651	29±1	25±1
T7351	31±1	30±1

The % RA loss in the peak-aged condition is of similar magnitude to that previously recorded with the commercial 7075 in the overaged condition.

This work will continue with 7000 alloys of other copper contents; other heat treatment conditions, particularly the underaged condition, will also be tested.

2.8 CONCLUSIONS

2.8.1 Hydrogen Embrittlement of Commercial 7075-T651

- The embrittling effect of cathodically-charged hydrogen has been confirmed.
- The embrittlement is manifested both as a loss in tensile ductility and as an increase in secondary cracking.

2.8.2 The Role of Microstructure in Hydrogen Embrittlement of Commercial 7075 Aℓ

- Ductility loss due to cathodically-charged hydrogen is a function of microstructure, being largest for the underaged microstructure and smallest for the overaged.
- This variation of the hydrogen effect with microstructure is strongly suggestive of a correlation with slip planarity, which in turn suggests an important influence of mobile dislocations on hydrogen transport.
- Intergranular fractures were not observed, owing to the orientation of the tensile axis parallel to the major grain boundary surfaces.

2.8.3 Strain Rate Dependence of Hydrogen Embrittlement in Commercial 7075 Aℓ

- Independent of microstructure, the hydrogen embrittlement exhibited a maximum at an intermediate strain rate.
- The reduced effect at low strain rates could be due to the diffusional escape of hydrogen from trap sites.
- At higher strain rates, the time of testing is too brief to allow the accumulation of hydrogen either through diffusion or dislocation transport, thus precluding it from influencing the fracture process.

2.8.4 Hydrogen Embrittlement of High Purity 7075 Aℓ

- The high purity alloy exhibited an intergranular fracture mode of hydrogen embrittlement.

- Dislocation transport of hydrogen, as accomplished by the SET procedure, controlled the extent and severity of the observed embrittlement.

2.8.5 Stress Corrosion Cracking of 7075: Mode I/Mode III Testing

- As has been previously demonstrated, the microstructural condition most susceptible to SCC is the underaged condition.
- Mode III loading, in which hydrogen's role in the fracture process should be minimized as there is effectively no triaxial stress state at the crack tip, resulted in much longer failure times at a given stress level.
- Mode I testing of notched longitudinal specimens confirmed the susceptibility hierarchy previously established: UT→T6→T73.
- Straining Electrode Testing of notched Mode I specimens confirmed the drop in fracture toughness of the underaged specimens.

2.8.6 Hydrogen Embrittlement of 2124 Aℓ

- It was demonstrated that, while it was less sensitive than 7075, 2124 could be embrittled by cathodically charged-hydrogen, provided that hydrogen entry conditions were severe.
- In general, the results conformed to the framework established for the 7075 alloys: the underaged microstructures were the ones susceptible to hydrogen embrittlement.

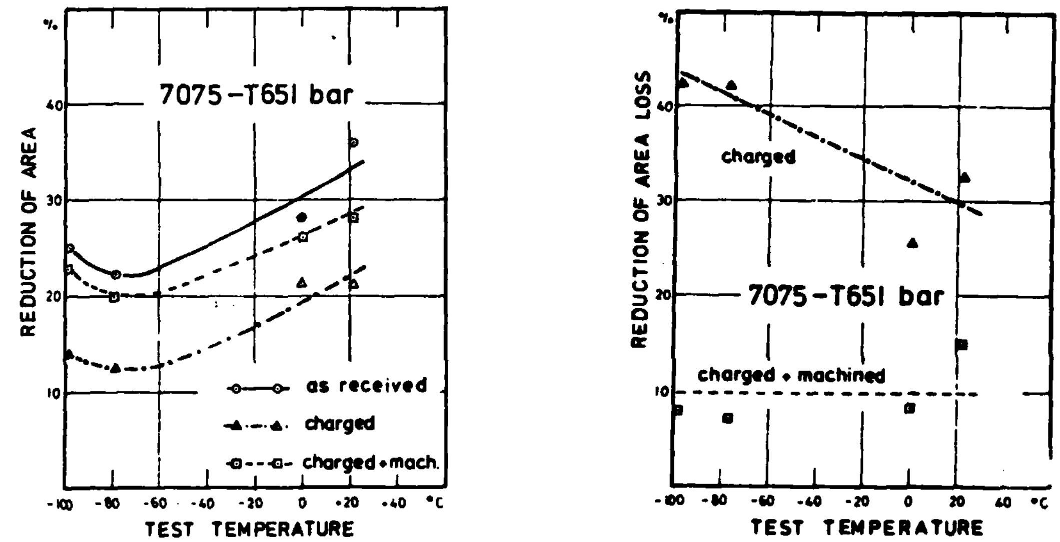


Figure 1: Ductility and ductility loss as functions of test temperature. (21)

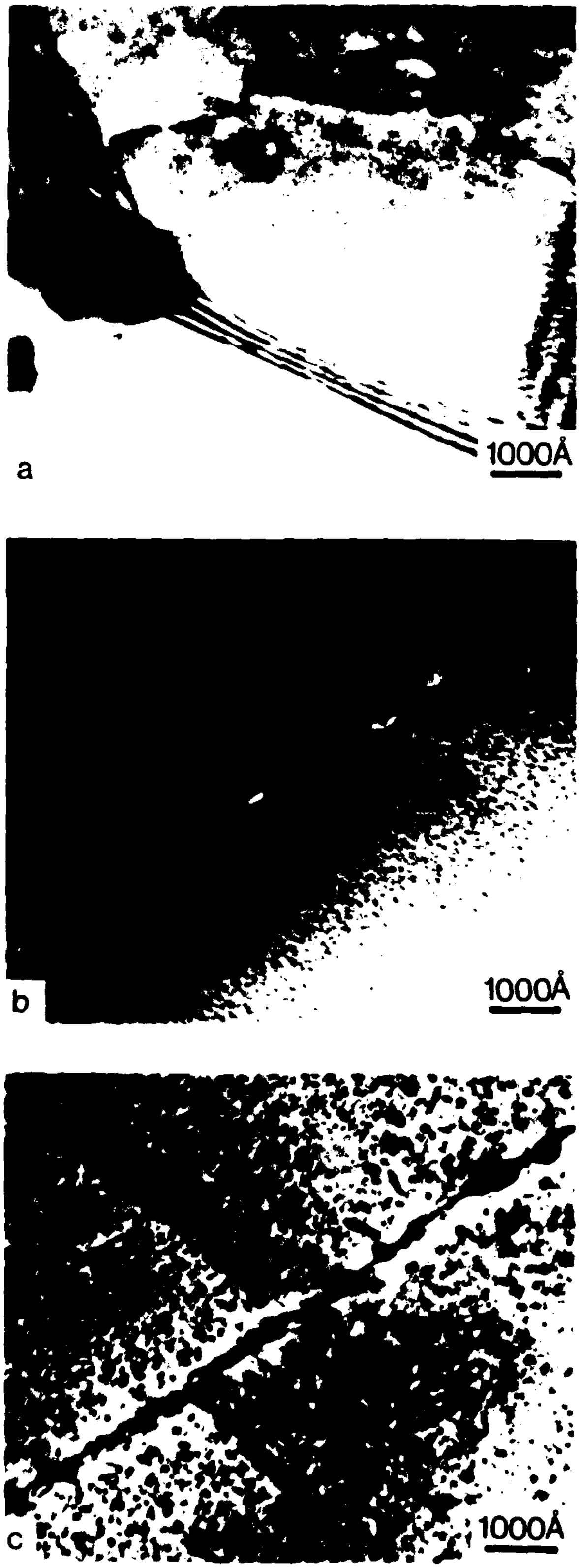


Figure 2: Transmission electron micrographs of microstructural details of the three heat treatments: (a) underaged, UT, (b) T6, (c) T73. (6)

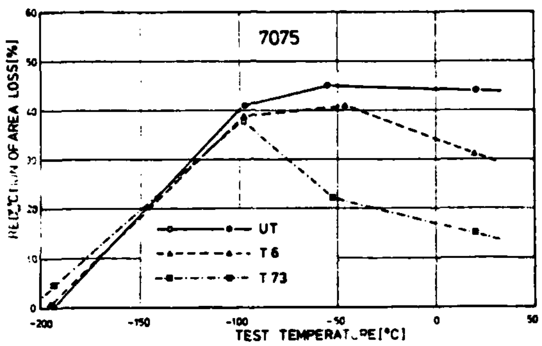


Figure 3: Reduction of area loss vs test temperature for the three different microstructures. (6)

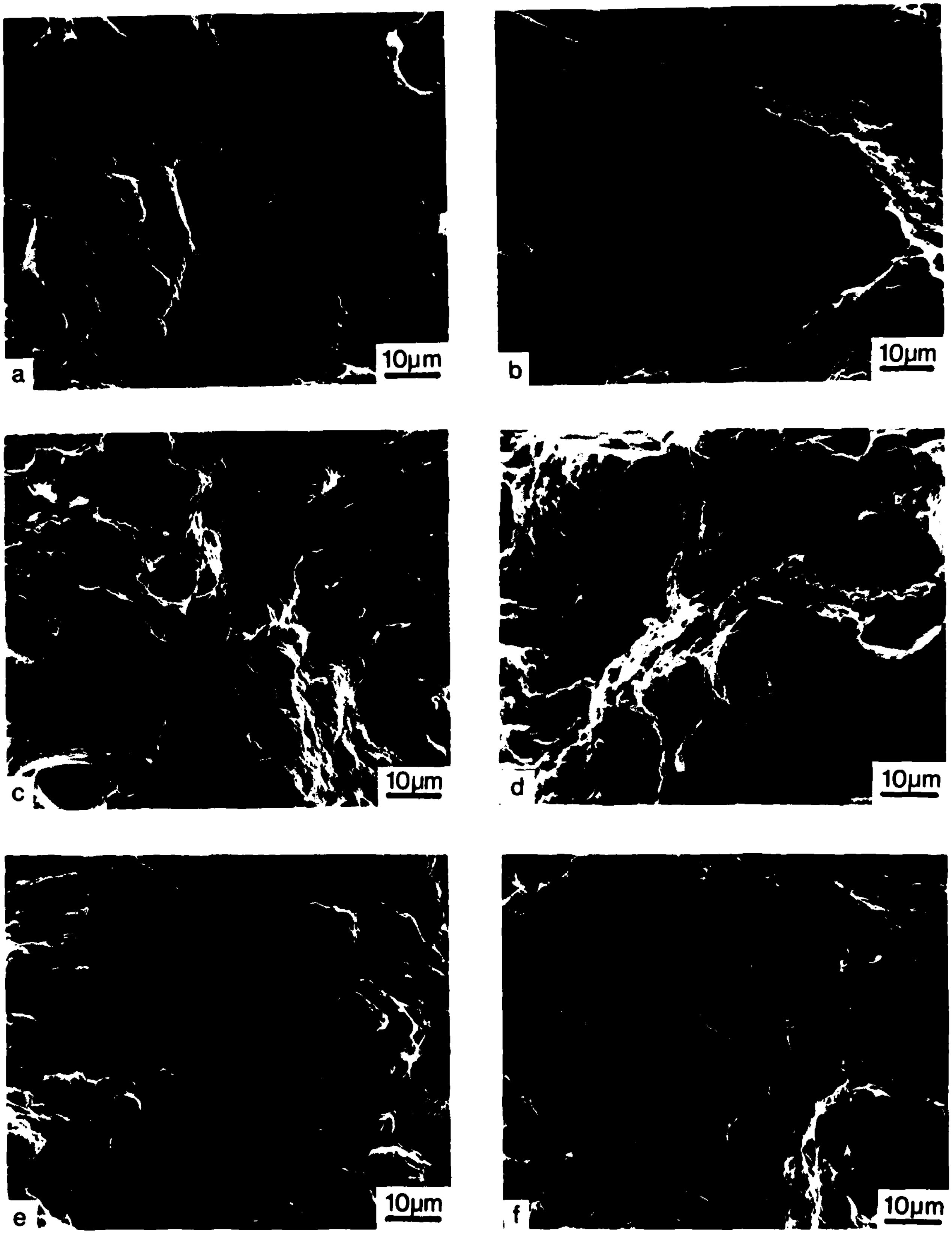


Figure 4: Typical fracture surfaces (SEM), test temperature 20°C: (a) underaged, uncharged, (b) underaged, charged, (c) T6, uncharged, (d) T6 charged, (e) T73, uncharged, (f) T73, charged. (6)

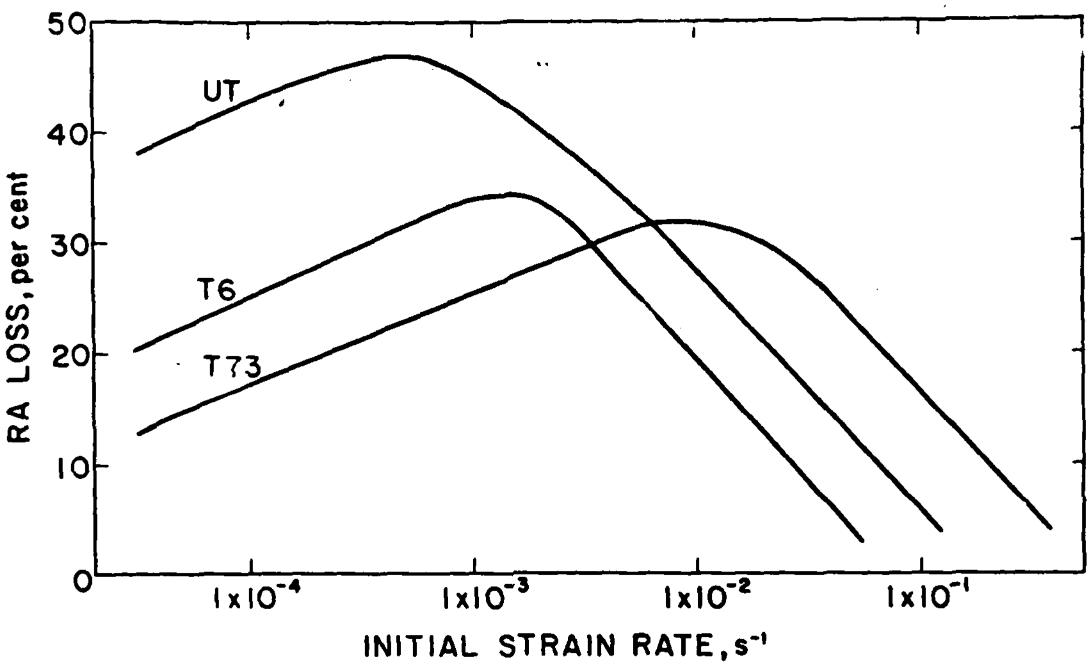


Figure 5: RA loss in hydrogen-charged commercial 7075 as a function of strain rate (7).

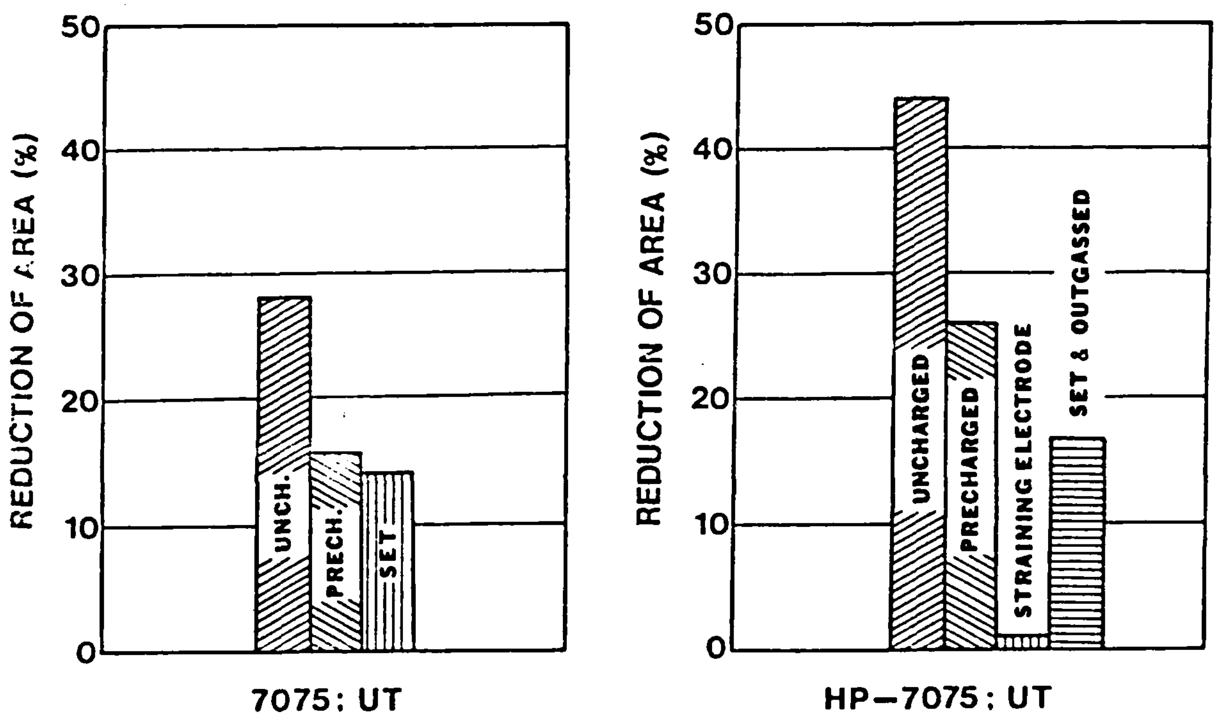


Figure 6: Effect of cathodic charging procedure, including straining electrode tests (SET), on hydrogen susceptibility. Left: Underaged (UT) commercial 7075. Right: UT high-purity 7075. (9)

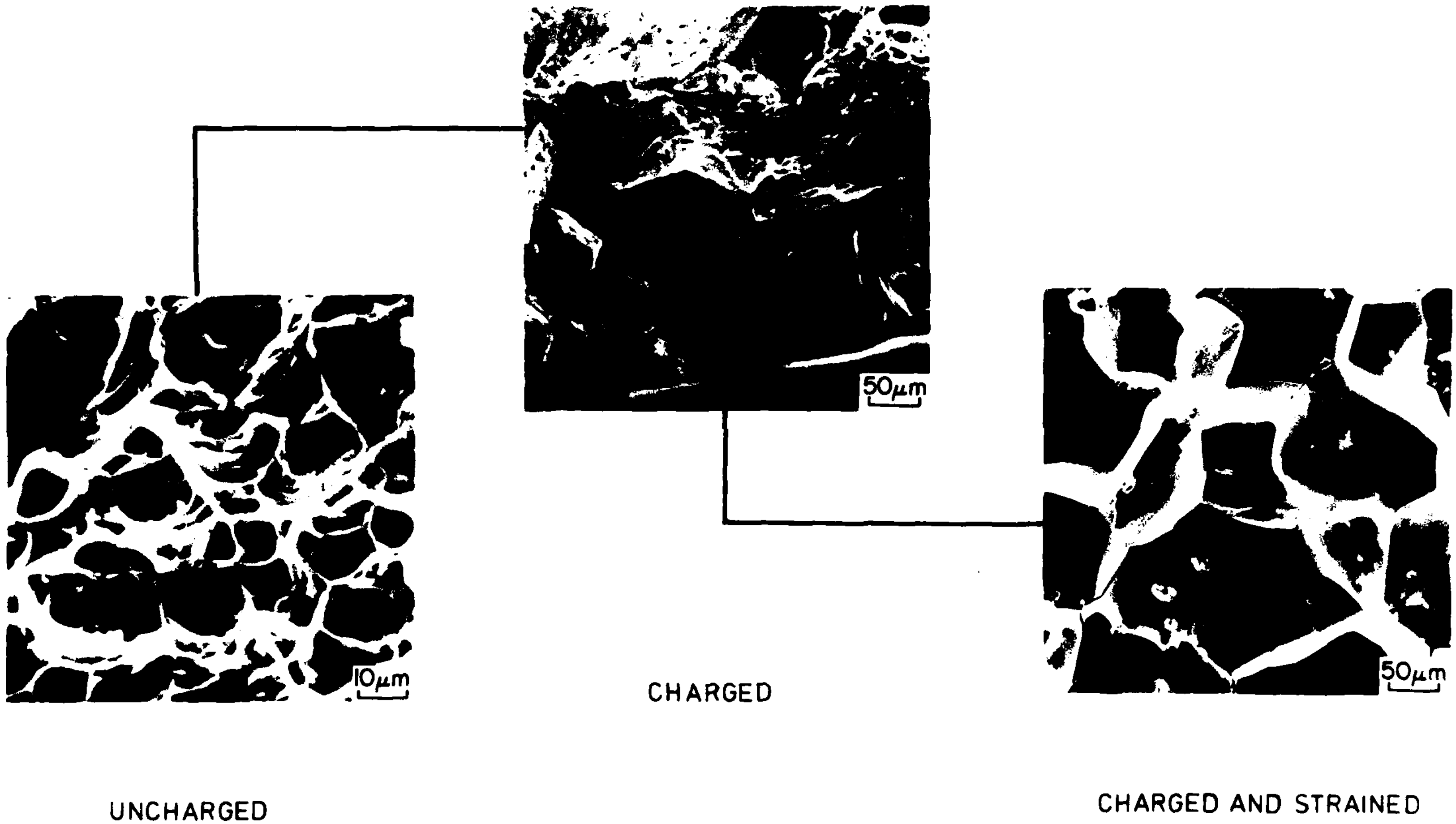


Figure 7: Comparison of hydrogen pre-charged (center) fractography of HP7075, with uncharged and SET specimens. (9)

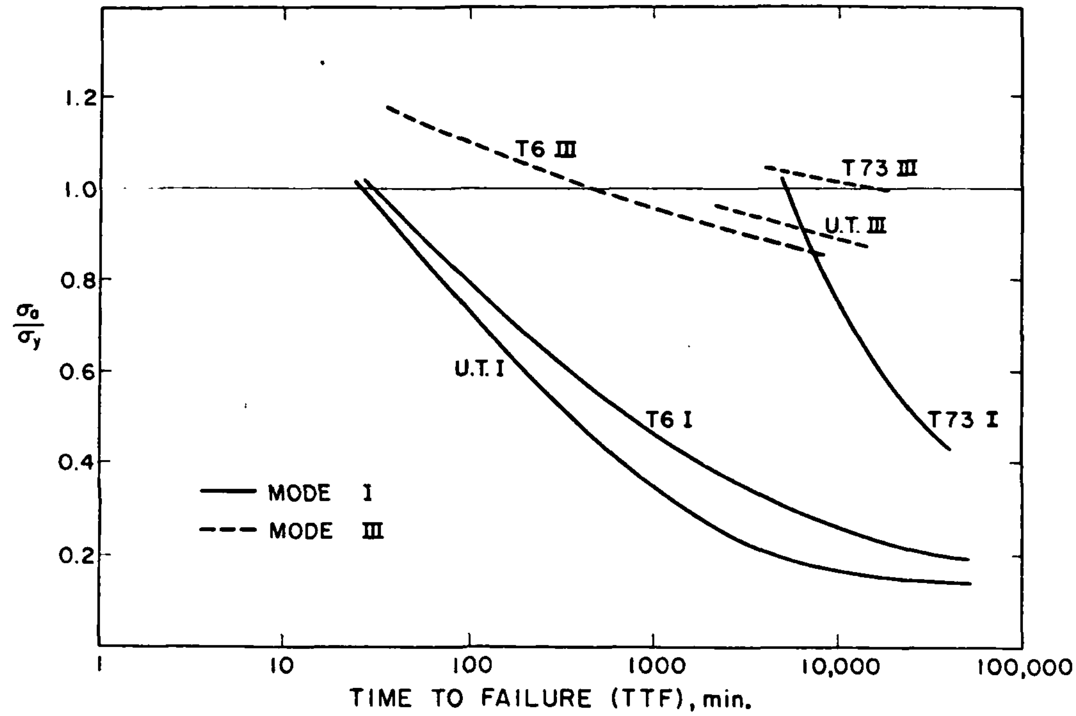


Figure 8: Stress corrosion cracking results for short transverse specimens. (12)

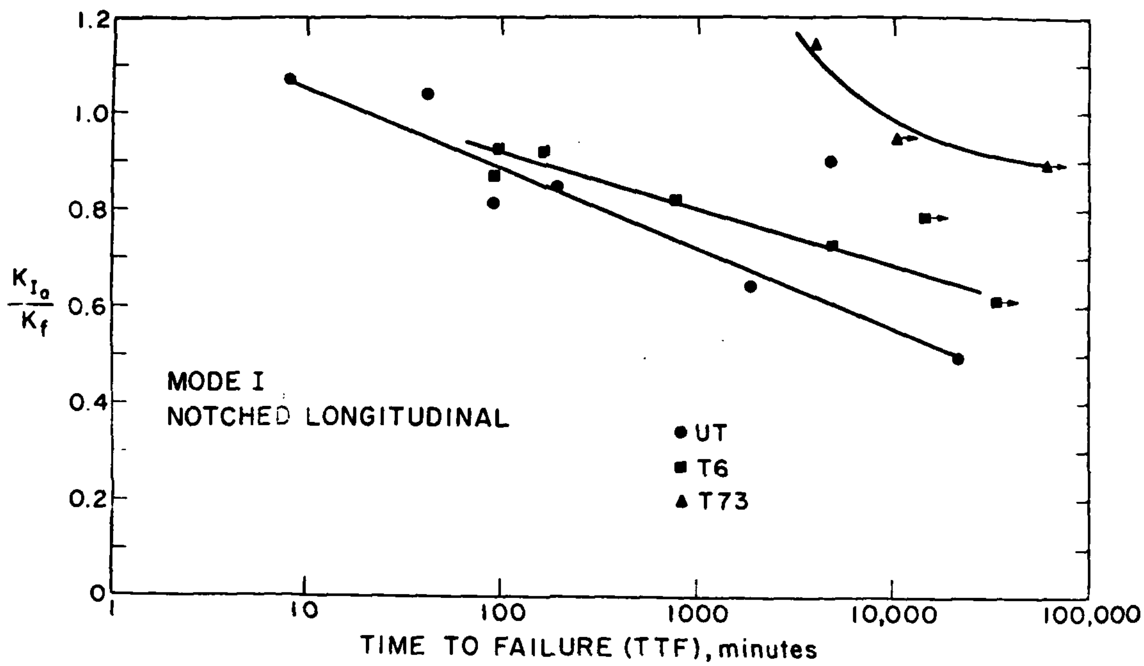


Figure 9: Stress corrosion cracking results for notched longitudinal specimens. (12)

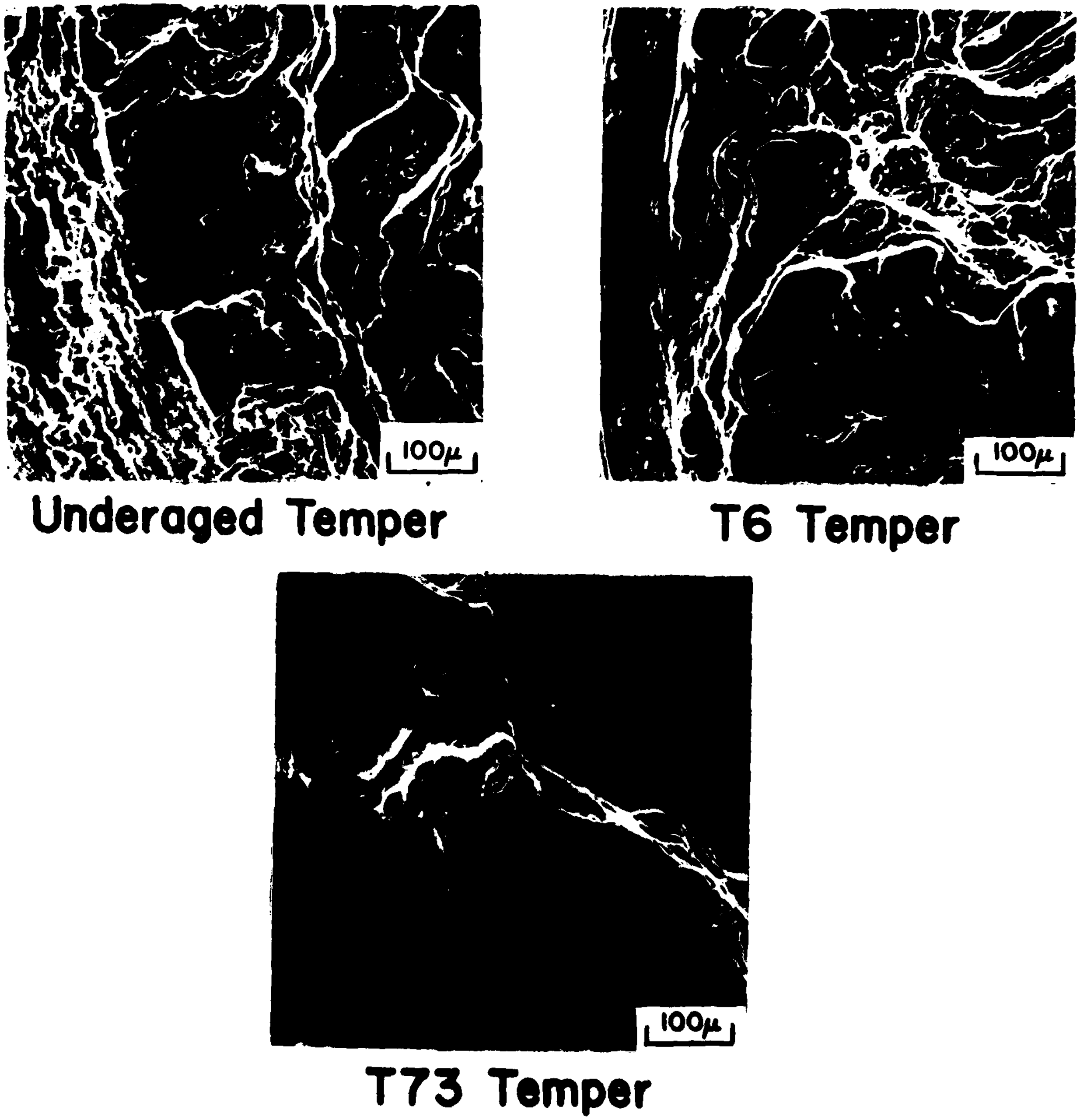
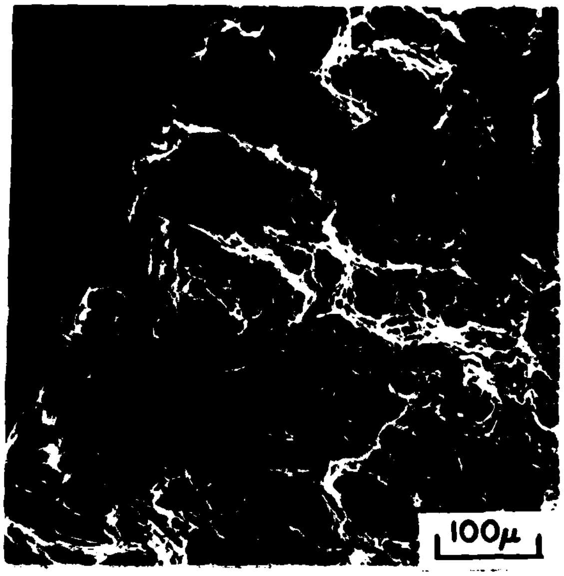
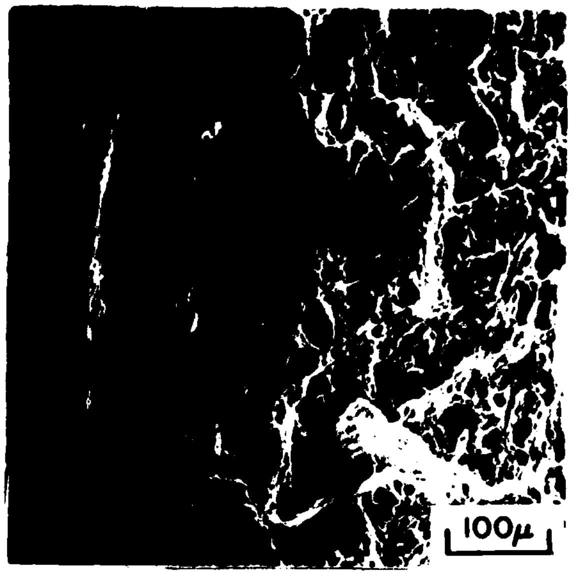


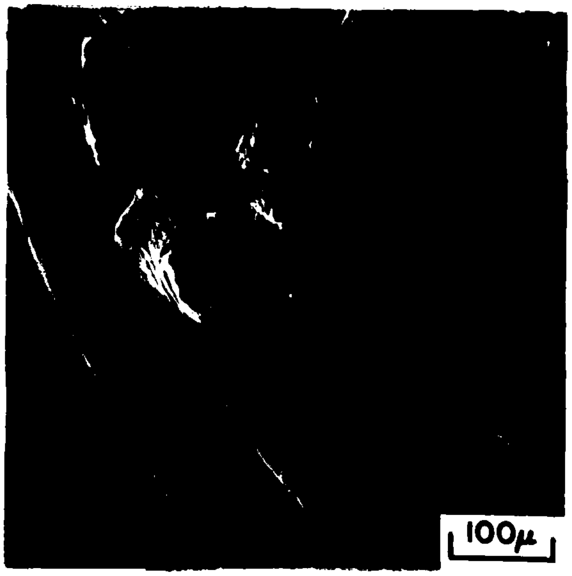
Figure 10: Fractography of short transverse tensile SCC specimens



Underaged Temper



T6 Temper



T73 Temper

Figure 11: Fractography of longitudinal tensile SCC specimens.

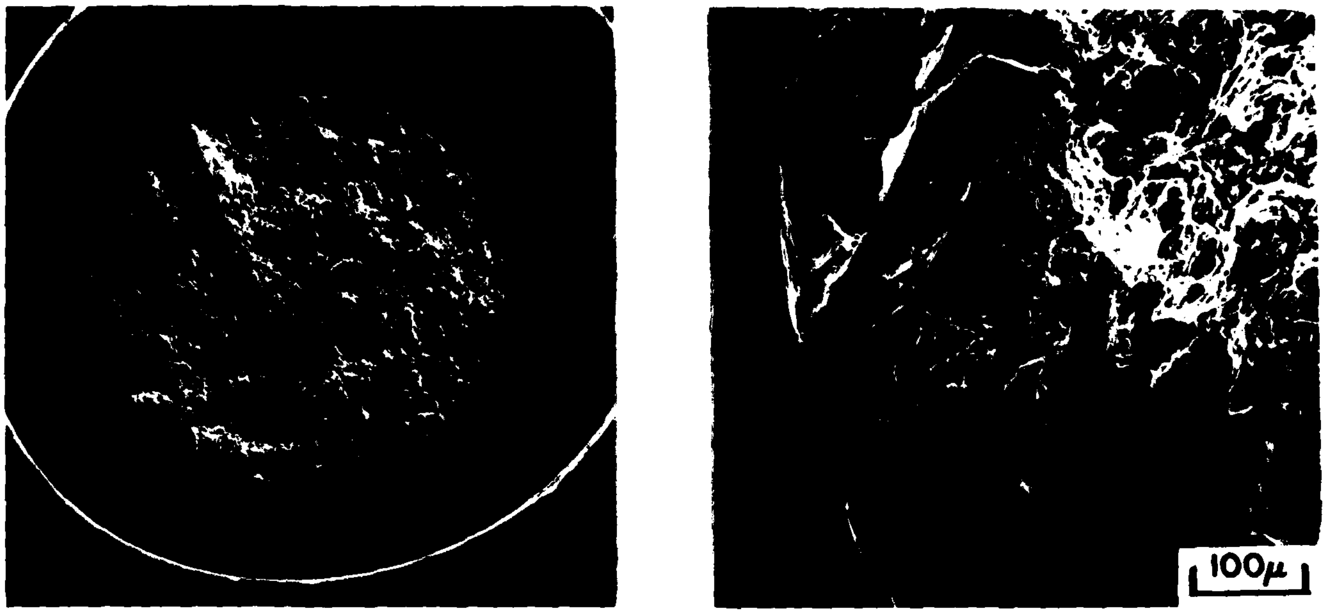


Figure 12: Fractography of longitudinal, underaged (UT) specimens following SET testing. (12)

2.9 References

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9. J. Albrecht, I. M. Bernstein and A. W. Thompson: "The Role of Hydrogen Transport by Dislocations in Hydrogen Embrittlement of Aluminum": Met. Trans. A, submitted.

10. J. A. S. Green, H. W. Hayden and H. Montague: Effect of Hydrogen on Behavior of Materials (A. W. Thompson and I. M. Bernstein, eds.), TMS-AIME, New York, 1976, pp. 200-15.

11. J. L. Maloney, A. W. Thompson and I. M. Bernstein: "Effect of Microstructure and Loading Mode on Stress Corrosion of 7075 Aluminum", in preparation.

12. R. E. Swanson, A. W. Thompson, I. M. Bernstein and J. L. Maloney: Hydrogen Effects in Metals (A. W. Thompson and I. M. Bernstein, eds.), TMS-AIME, in press.

13. A. W. Thompson: Mater. Sci. Eng., 1980, vol. 43, pp. 41-46.

14. D. A. Hardwick, M. Taheri, A. W. Thompson and I. M. Bernstein: "Hydrogen Embrittlement in a 2000-series Aluminum Alloy": Met. Trans. A, submitted.

3.0 PUBLICATIONS AND PRESENTATIONS

3.1 Publications

The following list includes all publications or papers in press which derived support from this grant. Those which also appear in the list of references, Section 2.9, are indicated with the reference number in parentheses.

1. A. W. Thompson and I. M. Bernstein, "The Role of Plastic Fracture Processes in Hydrogen Embrittlement", in Fracture 1977 (Vol. 2), Univ. Waterloo Press (1977), pp. 403-07.

2. I. M. Bernstein and A. W. Thompson, "An Evaluation of Hydrogen Embrittlement Mechanisms", in Proc. Conf. on Mechanisms of Environmental Embrittlement in Materials, Metals Society, London (1978), pp. 412-26.

3. A. W. Thompson and I. M. Bernstein, "Influence of Hydrogen in Plastic Fracture Processes", in Proc. Second Int. Congress on Hydrogen in Metals (Vol. 10), Pergamon Press (1977), paper 3A6.

4. J. Albrecht, B. J. McTiernan, I. M. Bernstein and A. W. Thompson, "Hydrogen Embrittlement in a High-Strength Aluminum Alloy", Scripta Met., vol. 11 (1977), pp. 893-97. (Ref. 2)

5. J. Albrecht, A. W. Thompson and I. M. Bernstein: "The Role of Microstructure in Hydrogen-Assisted Fracture of 7075 Aluminum", Met. Trans., vol. 10A (1979), pp. 1759-66. (Ref. 6)

6. M. Taheri, J. Albrecht, I. M. Bernstein and A. W. Thompson: "Strain-rate on Hydrogen Embrittlement of 7075 Aluminum", Scripta Met., vol. 13 (1979), pp. 871-75. (Ref. 7)

7. I. M. Bernstein and A. W. Thompson: "Effect of Structural Inhomogeneities on Hydrogen Embrittlement", in Hydrogen in Metals: Trans. Japan Inst. Metals, 21 (Suppl.) (1980), pp. 429-32.

8. A. W. Thompson and I. M. Bernstein: "Critical Tests of the Role of Hydrogen in Stress Corrosion Cracking", in Hydrogen in Metals: Trans. Japan Inst. Metals, 21 (Suppl.) (1980), pp. 589-92.

9. A. W. Thompson, "Current Status of the Role of Hydrogen in Stress Corrosion Cracking", Mater. Sci. Eng., vol. 43 (1980), pp. 41-46. (Ref. 13)

10. R. E. Swanson, A. W. Thompson, I. M. Bernstein and J. L. Maloney: "Effect of Stress State on the Stress Corrosion Cracking of 7075 Aluminum", in Hydrogen Effects in Metals (Proc. 3rd Int. Conf. on Hydrogen, Jackson, WY, August 1980), TMS-AIME, in press. (Ref. 12)

11. J. Albrecht, I. M. Bernstein and A. W. Thompson: "The Role of Dislocation Transport in the Hydrogen Embrittlement of 7075 Aluminum", Met. Trans. A, submitted. (Ref. 9)

12. D. Hardwick, M. Taheri, A. W. Thompson and I. M. Bernstein: "Hydrogen-assisted Fracture of a 2000-Series Aluminum Alloy", Met. Trans. A, submitted. (Ref. 14)

13. J. L. Maloney, A. W. Thompson and I. M. Bernstein: "Effect of Microstructure and Loading Mode on Stress Corrosion of 7075 Aluminum", in preparation. (Ref. 11)

14. D. A. Hardwick, A. W. Thompson and I. M. Bernstein: "Hydrogen-Assisted Fracture of 7050 Aluminum", in preparation.

3.2 Presentations

The following list includes conference presentations and other technical talks based wholly or in part on work done with the support of this grant; several of these correspond to the publications in Section 3.1, above, and are so indicated below. For multiple authors, speaker is underlined.

1. Fourth International Conference on Fracture, Waterloo, Ontario, June 1977 (see publication 1, above).

2. University of Surrey, U.K. (see Publication 2, above).

3. Second International Congress on Hydrogen in Metals, Paris, June 1977 (see publication 3, above).

4. TMS-AIME Fall meeting, Chicago, October 1977 (see publication 4, above).

5. A. W. Thompson, "Hydrogen Embrittlement and Stress Corrosion Cracking of Structural Materials", WESTEC Conference, Los Angeles, March 1977.

6. I. M. Bernstein, "Hydrogen Embrittlement", Columbia University, April 1977.

7. I. M. Bernstein, "Overview of Hydrogen Embrittlement Studies", Berkeley Nuclear Laboratories, U.K., September 1977.

8. A. W. Thompson, "Effect of Metallurgical Variables on Environmental Fracture of Engineering Materials", TMS-AIME Fall Meeting, Chicago, October 1977.

9. I. M. Bernstein, "Stress Corrosion Cracking and Hydrogen Embrittlement", Central Electricity Research Lab - Surrey, England, January 1978.

10. A. W. Thompson, "Advances in Understanding of Hydrogen Embrittlement", Materials Science Lecture Series, Westinghouse R & D Center, February, 1978.

11. A. W. Thompson, "Hydrogen Embrittlement in Iron-, Aluminum- and Titanium-base Alloys", Syracuse University, April 1978.

12. I. M. Bernstein, "Alloy Design and Hydrogen Embrittlement", The Technion-Haifa, Israel, April 1978.

13. I. M. Bernstein, "Hydrogen Embrittlement of Aluminum Alloys", Imperial College, London, July 1978.

14. J. Albrecht, A. W. Thompson and I. M. Bernstein, "The Influence of Microstructure on Hydrogen Embrittlement in 7075 Aluminum", Fall Meeting, TMS-AIME, St. Louis, October 1978 (see publication 5, above).

15. J. Albrecht, I. M. Bernstein and A. W. Thompson, "Role of Hydrogen in Embrittlement of 7075 Type Alloys", AIME Annual Meeting, New Orleans, LA, February 1979 (see publication 11, above).

16. I. M. Bernstein and A. W. Thompson, "Hydrogen Embrittlement of Aluminum Alloys", Imperial College, London, March 1979.

17. I. M. Bernstein and A. W. Thompson, "Role of Hydrogen in Stress Corrosion Cracking", NACE Annual Meeting, Atlanta, GA, March 1979.

18. I. M. Bernstein, "Relation Between Hydrogen Embrittlement and Stress Corrosion Cracking", ASM Short Course, Cleveland, OH, May 1979.

19. I. M. Bernstein, "Hydrogen Embrittlement in Structural Alloys", Brown University, Providence, RI, May 1979.

20. I. M. Bernstein and A. W. Thompson, 2nd Japan Inst. Metals Int. Symposium, on Hydrogen in Metals, Mirakami, Japan, Nov. 1979 (see publication 7, above).

21. A. W. Thompson and I. M. Bernstein, Ibid. (see publication 8, above).

22. J. L. Maloney, I. M. Bernstein and A. W. Thompson, "The Role of Microstructure and Loading Mode in Stress Corrosion Cracking of 7075 Aluminum", AIME Annual Mtg., Las Vegas, Feb. 1980. (see publication 13, above).

23. D. Hardwick, M. Taheri, A. W. Thompson and I. M. Bernstein, "Hydrogen Embrittlement in a 2000-series Aluminum Alloy", AIME Annual Mtg., Las Vegas, Feb. 1980 (see publication 12, above).

24. R. E. Swanson, I. M. Bernstein, A. W. Thompson and J. L. Maloney, "Effect of Stress State on the Stress Corrosion Cracking 1f 7075 Aluminum", 3rd Int. Conf. on Hydrogen in Materials, Jackson Lake Lodge, WY, Aug. 1980 (see publication 10, above).

25. D. A. Hardwick, A. W. Thompson and I. M. Bernstein, "The Influence of Microstructure on the Hydrogen-assisted Fracture of 7050 Aluminum", Fall Meeting, TMS-AIME, Pittsburgh, Oct. 1980 (see publication 14, above).

26. R. E. Swanson, I. M. Bernstein and A. W. Thompson, "Effect of Stress State on the Stress Corrosion Cracking of 7075 Aluminum in the T6-RR Temper", Fall Meeting, TMS-AIME, Pittsburgh, Oct. 1980.

4.0 PERSONNEL

A. W. Thompson - Professor and Co-principal Investigator (20% AY, 50% summer): 1 Dec. 76 - 31 Dec. 80.

I. M. Bernstein - Professor and Co-principal Investigator (5% AY, 30% summer): 1 Dec. 76 - 31 Dec. 80.

J. Albrecht - Post-doctoral Associate: 1 Feb. 77 - 15 Apr. 79.

D. Hardwick - Post-doctoral Associate: 1 Sept. 79 - 31 Dec. 80.

R. Lober - Graduate Student: 1 Sept. 77 - 1 May 78.

J. Maloney - Graduate Student: 15 June 78 - 30 June 80. (M.E. Degree)

R. E. Swanson - Graduate Student: 1 Sept. 79 - 31 Dec. 80 (Ph.D. candidate)

B. J. McTiernan - Undergraduate Assistant: 1 Dec. 76 - 1 June 77.

Presented below are brief current resumes of the two Principal Investigators on this project.

Anthony W. Thompson

Professor Thompson obtained his B.S. at Stanford University, spent a year at the Jet Propulsion Laboratory before receiving a M.S. at the University of Washington, and received his Ph.D. in 1970 at M.I.T. He then spent three years at Sandia Laboratories, Livermore, and four years at the Science Center of Rockwell International before joining Carnegie-Mellon University, where he is now Professor of Metallurgy and Materials Science. Dr. Thompson's interests encompass the relation of microstructure to a wide variety of mechanical properties, particularly deformation and fracture. Primary examples of these interests include investigation of many materials for hydrogen compatibility; several studies on fatigue, with emphasis on grain size effects; fundamental work on fracture mechanisms; and an analysis of polycrystal work hardening. He is a member of The Metallurgical Society of AIME, Sigma Xi, ASM, and the American Association for the Advancement of Science; he is active in TMS-AIME, serving as Chairman of the Physical Metallurgy Committee, Past Chairman of the Mechanical

Metallurgy Committee, and member of several other committees. He is currently an Associate Editor for Metallurgical Transactions and is a member of the editorial board of International Metals Reviews. He and I. M. Bernstein have organized and chaired three international conferences on hydrogen, held at Seven Springs, PA in 1973 and at Jackson, Wyoming in 1975 and 1980. Dr. Thompson has authored or co-authored 110 technical papers and over 170 technical presentations on his research.

I. M. Bernstein

Dr. Bernstein is Professor of Metallurgy and Materials Science at Carnegie-Mellon University and Associate Dean of Carnegie Institute of Technology. He obtained his B.S., M.S. and Ph.D. (1966) at Columbia University. After 18 months as a post-doctoral research fellow at the Berkeley Nuclear Laboratory in England, he joined the E. C. Bain Laboratory for Fundamental Research of the U. S. Steel Corporation, where he performed research on the fundamentals of deformation and fracture in iron. This included studies on the role of grain boundaries and grain boundary structure on mechanical properties; on environmental effects, particularly hydrogen embrittlement in iron and iron alloys; and on characterization of structure by electron microscopy. He joined Carnegie-Mellon University in 1972 where he has continued research in the above areas. He is a member of the Metallurgical Society of AIME, ASM, American Society for Engineering Education and the American Association for the Advancement of Science and is active in the technical affairs of several of these societies. He is an Associate Editor of Metallurgical Transactions, and of a Handbook of Stainless Steels (McGraw-Hill), and is coordinator of an ARPA sponsored project to produce two handbooks on stress corrosion cracking. He and A. W. Thompson have organized, chaired and are proceeding editors of three international conferences on hydrogen, held at Seven Springs, Pennsylvania in 1973 and at Jackson, Wyoming in 1975 and 1980. He spent the 1977-78 academic year at the London Office of Naval Research as their liaison scientist. He has authored or co-authored 55 technical papers and has presented more than 80 lectures on his research interests.

5.0 COUPLING

The principal investigators listed above have had extensive contact with other AFOSR investigators in related fields, as briefly summarized below.

Prof. W. W. Gerberich (AFOSR P.I. at University of Minnesota): Discussions on hydrogen accumulation at precipitates and grain boundaries in aluminum alloys. General discussions about aluminum alloy fracture modes and stress corrosion cracking phenomena.

Prof. J. L. Swedlow (AFOSR P.I. (solid mechanics) at Carnegie-Mellon University): Discussions of microstructural aspects of deformation and fracture in two-phase alloys.

Dr. Phillip Adler (Grumman Aerospace): Discussions on special techniques for measuring hydrogen contents, particularly localized distributions.

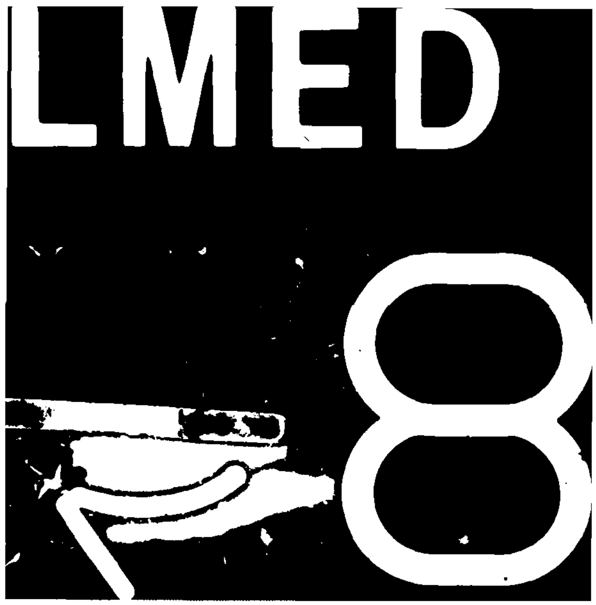
Prof. E. N. Pugh (formerly AFOSR P.I. at University of Illinois): Discussions about stress corrosion processes in aluminum alloys and hydrogen effects on fracture.

Prof. J. C. Williams (AFOSR P. I. at Carnegie-Mellon University): Discussions on strengthening mechanisms in aluminum alloys.

Prof. H. Margolin (AFOSR P.I. at Polytechnic Institute of New York): General discussions of the role of microstructure in fracture.

Dr. N. E. Paton (formerly AFOSR P.I. at Rockwell Science Center): Discussions about mechanical behavior and stress corrosion cracking of 7000-series aluminum alloys.

We also had the opportunity to meet with the entire AFOSR-sponsored hydrogen working group at the Milwaukee meeting of TMS-AIME in Sept., 1979. Also in attendance were other Air Force representatives such as from Wright Patterson. This provided a valuable opportunity for technical discussions of mutual interest.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